

Title: "Reduced Strain Silicon-Based Heterostructures for High Speed Electronic Devices"

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FINAL TECHNICAL REPORT

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Abstract

The main overall goal of this project was to improve the high speed performance of transistors and integrated circuits based on silicon, which is the material on which most microelectronic circuitry (such as microprocessors and memory chips) is based. More specifically, we addressed the three-fold gap (150 GHz vs. 50 GHz top speeds) between the performance of heterojunction bipolar transistors made using silicon-germanium technology in isolated devices in research labs vs. that achieved in integrated circuits in production. We have shown that the incorporation of small amounts of carbon into the silicon-germanium can overcome one of the main problems which lead to the much lower performance of integrated devices vs. that of isolated laboratory devices. The carbon is effective because it greatly reduces the unwanted movement of atoms during the microfabrication process.

Introduction to Technical Approach:

To improve the performance of silicon-based heterojunction devices, one must overcome two fundamental problems. The first is the strain which is present due to the large size of the germanium atom. This limits the thickness of $\text{Si}_{1-x}\text{Ge}_x$ layers which may be grown pseudomorphically without defects, and as more Ge is added to raise the bandgap offset with respect to silicon, this critical thickness becomes smaller. Second, one must control dopant profiles with nm precision to obtain optimum results. In practice is often more of a problem during the integration and fabrication of the device than during the growth of the original device profile. We address both of these issues through the use of $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ alloys. The small C atom compensates for the strain of the large Ge atoms, leading to increased critical thickness for fixed Ge fraction. Also, because the effect of C on the bandgap was established by us to be small, for a given bandgap $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ layers have a larger critical thickness than $\text{Si}_{1-x}\text{Ge}_x$ layers on Si. Second, it was found that the substitutional carbon in $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ layers reduces the diffusion of dopants in those layers and in nearby layers by gettering interstitial Si atoms. Thus this greatly improves the problem of dopant diffusion in devices as well.

Results and Impact/Applications:

Silicon-germanium is an "artificial material" which is grown in very thin layers which are measured on the scale of atomic dimensions in our lab and that of others (such as the Naval Research Laboratory) by techniques such as Chemical Vapor Deposition. When the layers are grown with the various atoms in the layers grown with the desired profiles controlled with near atomic precision, transistor performance can be much higher than that made without such layers. However, because relatively high temperatures are required during the subsequent micro-fabrication processes required to make transistors and IC's out of these layers, atoms in the structure can move and destroy the very precise profiles originally grown into the layers, thus lowering the transistor

performance. This effect in large part is responsible for the three-fold performance gap described above.

Device Impact of Small Amounts of Carbon in the HBT Base

We originally were adding carbon to $\text{Si}_{1-x}\text{Ge}_x$ base layers to reduce the strain by forming $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ alloys, as described later in this report. However, as part of this work, we have discovered that adding very small amounts of carbon to these layers reduces the undesired atomic motion by more than a factor of 10, a very fortuitous effect [1]. It occurs at carbon levels too low to have a substantial impact on the strain and bandgap, which was our original motivation, but the payoff is even larger. We demonstrated in our lab how this effect, even though it is not completely understood, can be used to make transistors whose performance is not degraded nearly as much by processing at high temperature that of ones made without carbon. Some luck was involved since it was found that the addition of carbon not only reduces the diffusion of atoms directly in the area near the carbon, but that this beneficial effect also extends some distance away from the region with the carbon [1-3]. Large amounts of carbon reduce diffusion the most, but the lifetime suffers if the carbon level is above 0.5%. However, because the effect extends some distance from the $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ layer, for large carbon concentrations the carbon was put in a region of the device where its detrimental effects were minimal, but still close enough to the region where atomic motion at high temperature is critical. The results were proved both by showing that the performance of transistors made using small amounts of carbon is better than those made without carbon, and also by directly measuring the motion of atoms (specifically the diffusion of boron).

Fig. 1 shows the performance of otherwise identical "heterojunction bipolar transistors based on silicon collector and emitter regions, but with a $\text{Si}_{0.8}\text{Ge}_{0.2}$ base. In one case, 0.5% C has been added to the base. Both device have been annealed at 647 °C to simulate the effect of a heat treatment during processing, which could cause the motion of atoms in the structure after it has been grown. These transistors are extremely sensitive to any motion of boron, which has been added to the base layers to provide the p-type electrical conductivity required in these layers. For proper device operation, the boron must stay completely within the SiGe region, and not extend into the silicon regions on either side. If the boron moves as little as a few nanometers (a few trillionths of a meter, a distance of about 20 atoms) into the silicon, the collector current (generally used as the output) of these devices falls rapidly [4,5]. This effect (specifically the formation of parasitic electrical barriers at the base-collector junction) can be partially circumvented by applying a reverse electrical bias to this base collector junction. Thus the collector current with no bias applied is less after the heat treatment than before, but if a base-collector bias is applied then the current rises back towards its original value. This explains the difference in the plots of collector current vs. collector-emitter bias (V_{CE}), where it is seen that in the devices without C the current highly depends on this collector voltage. Such a dependence of collector current on base-collector bias is a "signature" of this parasitic effect, which negates the effect of the heterojunction. In practice, it is desired to have transistors whose performance have the advantages of the heterojunction (primarily higher speed) and which do not have a collector current which depends on

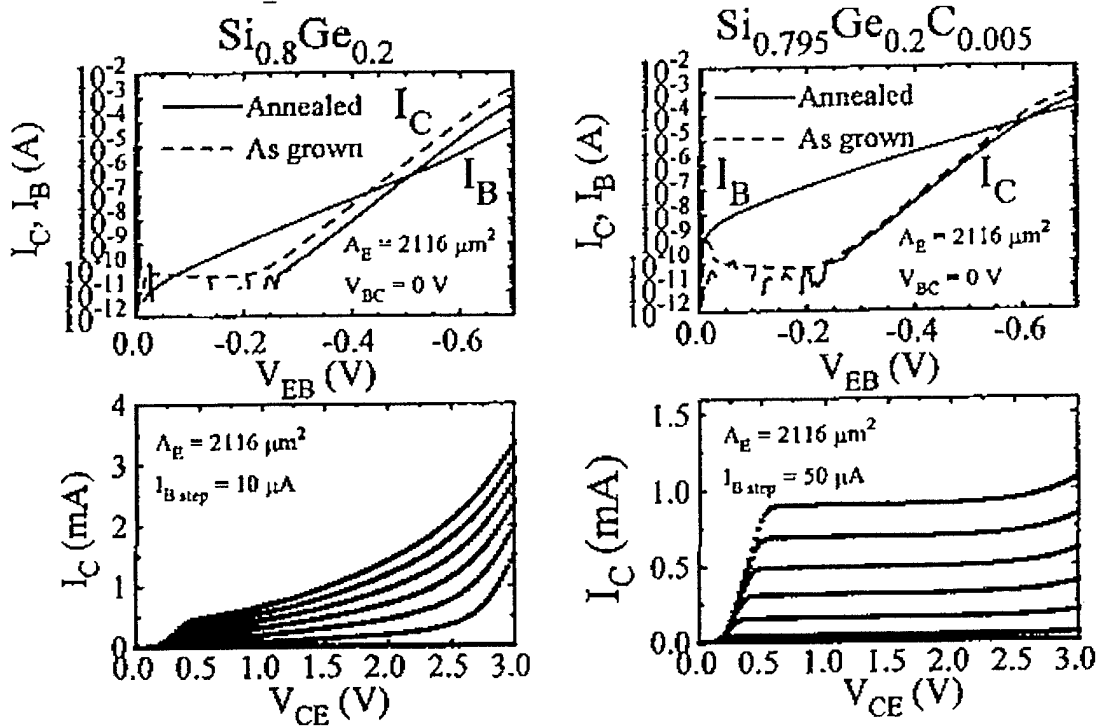


Figure 1. Collector current vs. base-emitter voltage (Gummel plots) and collector current vs. collector-emitter voltage (common-emitter characteristics) for devices after annealing at 647 °C for 15 minutes for transistors without and with 0.5% carbon in the base. (Ref. 1,2)

base-collector (or collector-emitter) bias. In Fig. 1, it is clear both in the plots of collector current vs. base-emitter voltage (Gummel plots) and in the plots of collector current vs. collector-emitter voltage that the effects of the transistor without C are severely degraded while those with C are not. The characteristics of the two transistors were nearly identical before annealing, and the annealing had little effect on the transistor with C. In the data shown here, the carbon does degrade the lifetime (a property of the carriers in the semiconductor), but by placing the carbon in the right place (not shown here for brevity) we are able to get around this effect by clever engineering of the location of the carbon relative to that of the boron and germanium in the base region of the transistor.

Fig. 2. shows the results of directly measuring the location of the boron atoms, showing that the improved transistor characteristics with carbon are due to reduced boron motion. The location of atomic concentrations vs. depth from the sample surface was measured with the Secondary Ion Mass Spectroscopy (SIMS) technique. The results are shown after annealing 755 °C (higher than the transistor results) to exaggerate the motion of the boron, since the SIMS technique, while the best available for the direct detection of atomic motion, is not nearly as sensitive to the electrical performance of the transistors themselves. Note that in the base regions there exists both boron and germanium, and that in one transistor a small amount of carbon has been added. The thickness of the base

regions is about 20 nm. In the sample without carbon, after annealing the boron profile now has “tails” extending into the silicon regions without germanium (the emitter and collector regions). Before annealing, the boron profile was sharp without tails, as was the profile in the transistor with carbon was both before and after annealing. Thus the ability

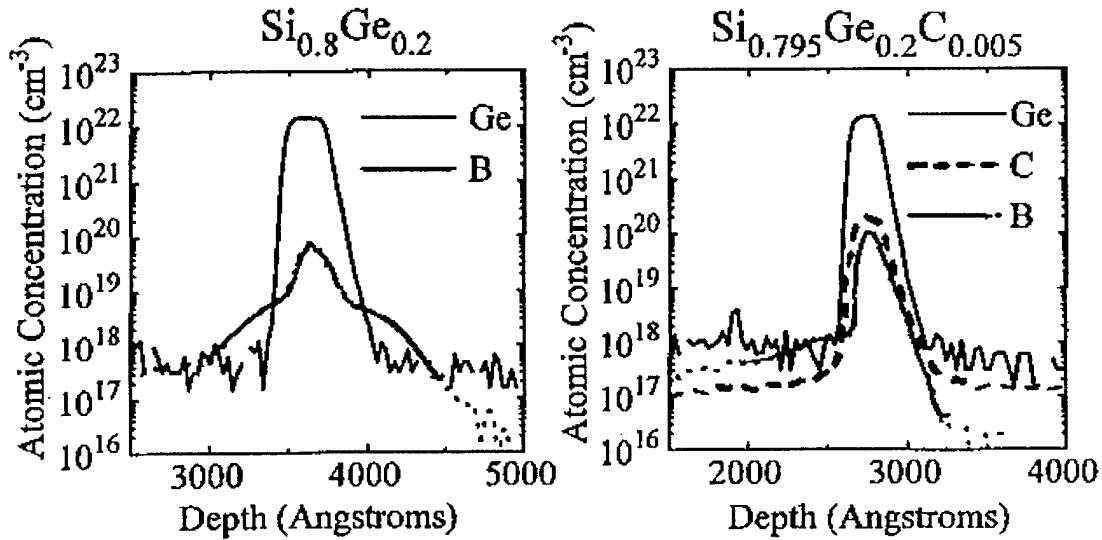


Figure 2. SIMS depth profiles of Ge, B, and C concentrations after annealing at 755 °C for 15 minutes in samples with and without carbon in the base layers. Note the difference in the boron profile in the sample with carbon. (Ref. 1,2)

of small amounts of carbon which in this case have a negligible effect on band offsets (< 10 meV) to drastically affect diffusion is clear.

Strain Reduction and Critical Thickness by adding C to $\text{Si}_{1-x}\text{Ge}_x$

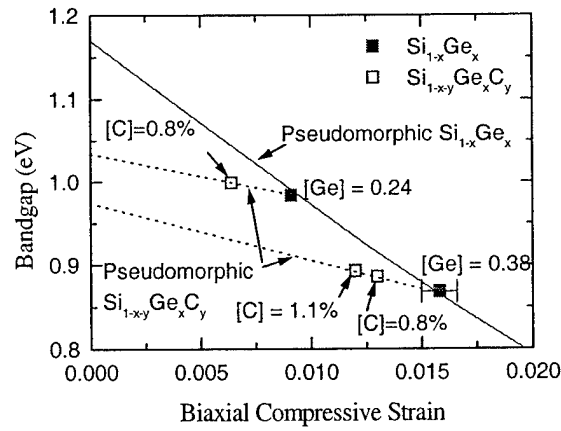


Figure 3. Bandgap vs biaxial compressive strain for pseudomorphic $\text{Si}_{1-x}\text{Ge}_x$ and $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ layers on Si(100) substrates [8].

Our original motivation for investigating carbon was the well-known limitations on strained $\text{Si}_{1-x}\text{Ge}_x/\text{Si}$ (100) heterojunction engineering by the strain in pseudomorphic layer. For example, to achieve a 100 meV bandgap reduction to Si, a Ge fraction of ~ 0.13 is required, with an equilibrium critical thickness of only ~ 25 nm (Fig. 3). In the early growth of $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$, it was established that the small size of the carbon atom leads to a reduction in the tensile strain of pseudomorphic $\text{Si}_{1-x}\text{Ge}_x$ layers on Si(100) substrates as C is added to form $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ layers [6], with each carbon atom compensating for the strain of 8-10 germanium atoms. Several years later as the quality of the materials improved, the effect of carbon on the bandgap was measured. As carbon is added to $\text{Si}_{1-x}\text{Ge}_x$ layers under compressive strain on Si(100), the bandgap increases as carbon is added at a rate of ~ 25 meV per percent of carbon [7-9], so that adding one carbon atom has the same effect on bandgap as removing three germanium atoms. This relatively slow rate of increase of bandgap as strain is reduced through the addition of carbon leads to lower strain and thus a higher critical thickness for a given bandgap for $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ compared to that for $\text{Si}_{1-x}\text{Ge}_x$ (Fig. 3, Fig. 4) [8,10]. This implies the bandgap for strain-free alloys lattice-matched to silicon is significantly less than that of silicon, which has been confirmed [11]. Adding 1% of C (and slightly increasing the Ge fraction) for our above target of a 100 meV bandgap change with respect to Si, a $\text{Si}_{0.83}\text{Ge}_{0.16}\text{C}_{0.01}$ layer would have a critical thickness of ~ 70 nm vs the 25 nm without carbon. If the carbon fraction were increased to a level of 0.02, the critical thickness of the $\text{Si}_{0.79}\text{Ge}_{0.19}\text{C}_{0.02}$ layer, also with a bandgap shift from Si of 100 meV, would be nearly one micron [10].

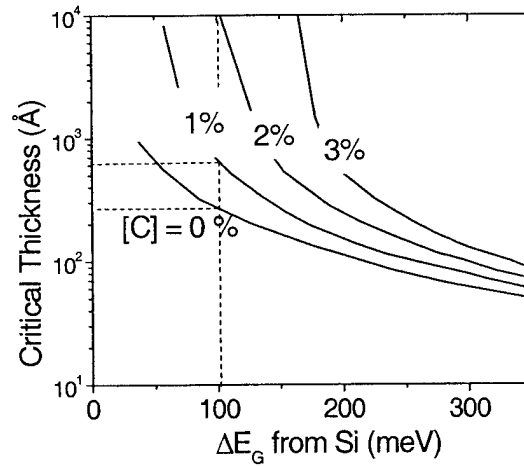


Figure 4. Critical thickness vs bandgap shift from silicon for compressively strained pseudomorphic $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ layers on Si (100) for 0, 1% and 2% carbon [10].

Fundamental Interaction of Carbon on Boron Diffusion through Silicon Interstitials

i-Si, 1500Å, 700°C
Si : boron $3 \times 10^{19}/\text{cm}^3$, 250Å, 700°C
i-Si, 100Å, 700°C
i- $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$, 250Å, 625°C
i-Si, 400Å, 700°C
Si : boron $3 \times 10^{19}/\text{cm}^3$, 250Å, 700°C
i-Si, 1500Å, 700°C
Si buffer layer : boron $3 \times 10^{19}/\text{cm}^3$, 2500Å, 1000°C

Fig.5. Test structure to study boron oxidation enhanced diffusion and transient enhanced diffusion. Two boron markers are separated by a $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ interstitial sink layer ($x = [0 - 0.2]$; $y = [0 - 0.005]$) (Ref. 3)

We used the structure of Fig. 5 to probe the mechanism of the reduction of diffusion. The structure has two marker layers of highly-boron-doped silicon layers designed to measure the reduction of oxidation enhanced diffusion of boron (OED) [3]. OED is caused by the injection of interstitial Si atoms, which are necessary for the diffusion of boron. One boron layer is above and one is below a $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ region designed to capture interstitials, which are introduced from the surface during oxidation. While the exact mechanism is not well understood, it appears that substitutional C creates a sink for Si interstitials, which are required for the diffusion of dopant atoms such as

boron. By reducing the local interstitial concentration, diffusion coefficients can thus be reduced, not just in the SiGeC region but also in nearby silicon regions. Note in Fig. 6 that in the control sample without any SiGeC (Fig 6(a.)) there is a substantial increase in the diffusion coefficient when the samples are annealed in oxygen as opposed to in nitrogen. Modelling the SIMS curves show an increase of the diffusion coefficient by nearly an order of magnitude. In the sample of Fig. 6(b.), a $\text{Si}_{0.8}\text{Ge}_{0.2}\text{C}_{0.005}$ layer was inserted between the boron marker layers. In this sample no OED was observed for the lower boron peak, showing that all of the injected interstitials were effectively captured. Transient-enhanced-diffusion (TED) in silicon is caused by the injection of interstitials resulting from implantation damage, and can cause a similarly large anomalous increase in boron diffusion. Because TED relies on an interstitial mechanism, it can similarly be avoided using SiGeC layers to block interstitials [3].

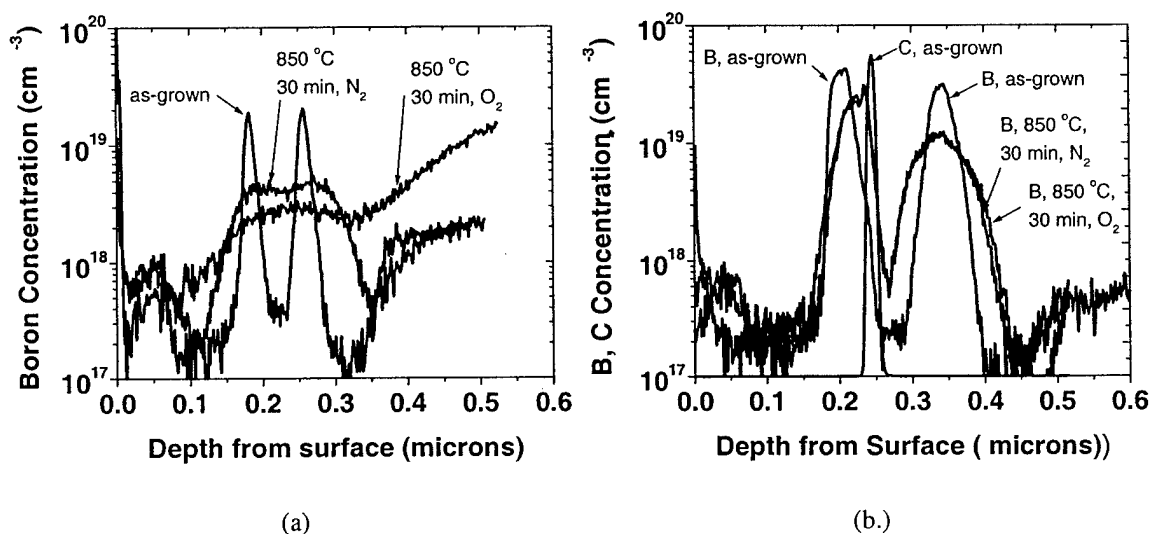


Figure 3. SIMS of boron and carbon levels (C as-grown only) in buried marker layers as-grown and after annealing at 850 °C in either N_2 or O_2 in (a.) control sample with no SiGeC and (b.) sample with SiGeC layer between the two boron layers. Note the elimination of OED in the lower boron-doped layer in sample (b.) [3].

Ongoing Challenges

There are two sets of ongoing challenges. The first involve quantifying the exact interaction between interstitial carbon and silicon, and thus determining the exact mechanism. How much carbon interacts with how many interstitials? This will require determination of the exact atomic level interaction between the interstitial silicon and substitutional carbon atom. A related question is what the electrical impact of any product, such as a precipitate, might be.

The second set of unknown is commercial. How can we move the technology to have positive impact for the Navy? It clearly appears from this work that their will be

clear applications for SiGeC, and that it probably can be implemented now, before the fundamental challenges mentioned above are solved. Several providers of SiGe HBT's, such as Motorola and Northrup Grumman, are at present pursuing this technology as a result of this work, but further support may be necessary. However, after many years of trying to reduce background contamination (such as carbon) in silicon based epi, it may take some time to convince production people to reintroduce carbon into epitaxial growth systems used for multiple purposes, because of fears of cross-contamination.

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Technology Transition/Students Supported

The Ph.D. student who has done a large part of this work, Louis Lanzerotti, has recently accepted an employment offer from IBM in Essex Junction, Vermont to work on the commercialization and manufacturing of such silicon-germanium semiconductor technology. A second Ph.D. student who was funded on the grant, Dr. Chialin Chang, now works at Motorola in Austin, Texas.

Motorola is now actively pursuing this technology for production, and expects to have initial products available in 2000. IBM has a large existing SiGe/Si HBT project, and is now beginning to investigate SiGeC as well to overcome their long-range problems. Northrup Grumman, which currently supplies the Navy with radar power sources with SiGe/Si HBT's, is also developing the technology.

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Suppression of boron transient enhanced diffusion in SiGe heterojunction bipolar transistors by carbon incorporation

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In this work, we demonstrate that the incorporation of carbon in the base of a *npn* Si/SiGe/Si heterojunction bipolar transistor dramatically reduces the outdiffusion of boron from the base under postgrowth implantation and annealing procedures. Without the addition of C, these processes would lead to transistors with vastly degraded transistor characteristics. This reduction in B diffusion, when compared to devices without C, has been observed by both secondary ion mass spectroscopy and improved electrical characteristics. © 1997 American Institute of Physics. [S0003-6951(97)00223-4]

Record high-frequency *npn* Si/Si_{1-x}Ge_x/Si bipolar transistor performance has been achieved by reducing the base sheet resistance using base boron doping levels greater than 10^{20} cm^{-3} .¹ However, high boron concentrations are susceptible to outdiffusion from the SiGe base into the Si emitter and collector following postgrowth thermal processing or implantation and annealing due to transient enhanced diffusion (TED) effects. This outdiffusion causes the formation of conduction-band barriers at the emitter-base and base-collector interfaces, which reduce the transistor's gain, early voltage, and frequency performance.²⁻⁵ To accommodate any boron movement during postgrowth processing, undoped SiGe spacer layers are grown on either side of the doped SiGe base.³ However, the thickness of these undoped spacer layers is limited by the critical thickness of the SiGe strained film. The necessity of maintaining low thermal budgets and the absence of implantation and annealing to minimize boron diffusion in heterojunction bipolar transistor (HBT) processes pose a severe limit on the integration of SiGe into base line silicon technology.

Boron outdiffusion may be caused by (1) thermal annealing,³ (2) transient enhanced diffusion effects due to an arsenic emitter implantation and anneal,² or (3) TED due to an extrinsic boron implantation and anneal.⁶ We have found that the addition of carbon to the base of SiGe HBTs sharply reduces boron outdiffusion and improves collector electrical characteristics in all cases. In this letter, we present results on TED due to an As emitter implant.

The *npn* Si/SiGeC/Si device epilayers were grown by rapid thermal chemical vapor deposition on (100) *p*-Si substrates.⁷ Following an *n*-type collector buffer layer, the 2000 Å $5 \times 10^{17} - 10^{18} \text{ cm}^{-3}$ collector was grown at 1000 °C. The 20% SiGe base was then grown at 625 °C, doped 10^{20} cm^{-3} boron, with nominally undoped spacer layers on either side of the doped SiGe base to compensate for any boron movement during the emitter growth. Identical wafers were grown using methylsilane to incorporate 0.5 at. % carbon in the doped SiGe, as well as in the undoped SiGe spacer layers. The emitter was then grown at 700 °C for 73 min, doped 10^{19} cm^{-3} with phosphorus. Transmission

electron microscopy (TEM) showed no dislocations, defects, or SiC precipitates in any of the as-grown layers. Figures 1(a) and 1(b) show secondary ion mass spectroscopy (SIMS) profiles of Ge, B, and C levels in the base region of as-grown layers without and with carbon, respectively, which demonstrate that the boron is contained well within the SiGe (or SiGeC) layers by the undoped spacer layers. Base oxygen levels are $\sim 2 \times 10^{18} \text{ cm}^{-3}$ in all cases. Transistors were processed from the as-grown wafers using a simple double mesa process in which the emitter metal is used as a mask for a selective wet etch, which etches the Si emitter but stops on the SiGe base. The collector is revealed by a plasma etch and, subsequently, the base and collector contacts are formed by lift-off. This process, which includes no high-temperature steps, is advantageous for this study because it prevents the base doping from moving during transistor fabrication. While the transistors have the ideal collector currents required for this study, they do suffer from nonideal base currents due to the lack of surface passivation.

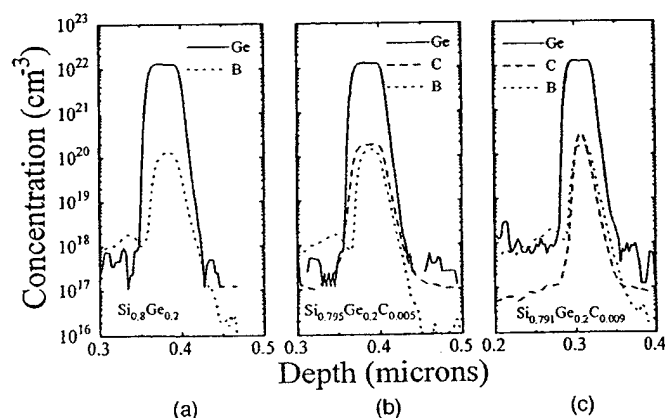


FIG. 1. SIMS profiles showing Ge, C, and B in as-grown base regions. Nominal structures are (a) *n*-Si/50 Å undoped SiGe spacer/200 Å SiGe: B-doped base/50 Å undoped SiGe spacer/*n*-Si, (b) *n*-Si/50 Å undoped SiGeC spacer/200 Å SiGeC: B-doped base/50 Å undoped SiGeC spacer/*n*-Si, and (c) *n*-Si/50 Å undoped SiGe spacer/50 Å SiGe: B-doped base/100 Å SiGeC(0.9% C): B-doped base/50 Å SiGe: B-doped base/50 Å undoped SiGe spacer/*n*-Si.

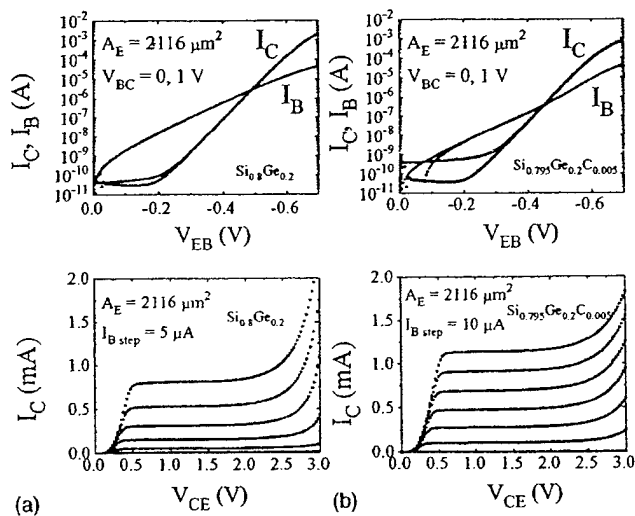


FIG. 2. Room-temperature Gummel plots and common-emitter characteristics of devices processed from Figs. 1(a) and 1(b) wafers without implantation or thermal annealing.

Previous work on SiGeC HBTs has shown that the addition of carbon to SiGe to form SiGeC alloys increases the band gap by ~ 26 meV/%C.⁸ Figures 2(a) and 2(b) show the Gummel plots and common emitter characteristics for the as-grown $\text{Si}_{0.8}\text{Ge}_{0.2}$ and $\text{Si}_{0.795}\text{Ge}_{0.2}\text{C}_{0.005}$ HBTs, respectively. The collector currents for both types of transistors have no observable dependence on base-collector reverse bias, which is reflected in the common-emitter characteristics by the high Early voltages. Since an increase in collector current with increased V_{CB} indicates a decrease in the parasitic barrier limiting I_C [evidence that outdiffusion has occurred], these electrical characteristics confirm the SIMS measurements that the as-grown transistors have no conduction-band barriers.

Pieces of the as-grown wafers with and without carbon were blanket implanted with As ($1.5 \times 10^{15} \text{ cm}^{-2}$, 30 keV; $3 \times 10^{14} \text{ cm}^{-2}$, 15 keV; chosen to follow Ref. 9) to form the emitter contact. Arsenic emitter implantation and annealing have been previously shown² to enhance boron diffusion in SiGe bases even though the As implantation range ($\sim 1000 \text{ \AA}$) is less than the emitter thickness (3000 \AA). Different implanted pieces were annealed at different temperatures ranging from 647 to 742 °C for 15 min in nitrogen, and, subsequently, double mesa transistors were fabricated on the as-annealed pieces. Figures 3(a) and 3(b) show Gummel plots and common-emitter characteristics of transistors following As implantation and annealing at 647 °C for wafers without and with carbon. The decrease in I_C and reduced Early voltage in the transistors without carbon show that boron has outdiffused even though this annealing condition is far less than the emitter thermal budget. Samples, which were not subjected to the As implant but underwent the same thermal annealing cycle, did not show any evidence of boron diffusion. This confirms that TED effects are responsible for the boron movement. However, the high Early voltages for the transistors with carbon illustrate that carbon in SiGe has suppressed the TED effects of an As emitter implant. Electrical results for 742 °C anneals are consistent with those at 647 °C, although the degradation due to diffusion in the de-

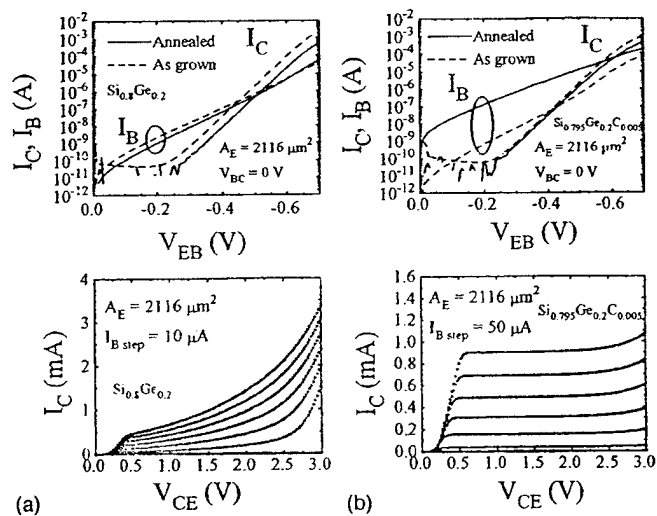


FIG. 3. Room-temperature Gummel plots and common-emitter characteristics of Figs. 1(a) and 1(b) wafers following As-emitter implantation and anneal at 647 °C.

vice without carbon is even more severe. Figure 4(a) and 4(b) show SIMS from no carbon and carbon wafers that underwent As implantation and a 755 °C, 15 min N_2 anneal. It is readily apparent that B has outdiffused in the transistor without carbon but has not in the transistor with carbon.

At both 647 and 742 °C, the samples with carbon showed large increases in base currents [Fig. 3(b)], indicative of the formation of deep level defects and a reduced lifetime in the base-emitter space-charge region. Since this did not occur in the devices without carbon, we conclude that this defect depends on carbon. It was confirmed that the excess base current was caused by recombination in the base-emitter depletion region by growing another sample where the B-doped SiGeC was sandwiched between SiGe layers (*without carbon*) of similar doping, which in turn were surrounded by the undoped SiGe (*without carbon*) spacers, as shown in Fig. 1(c). This prevents the penetration of the depletion regions into the SiGeC regions. These devices had only slight increases in base currents upon As implantation and annealing, as shown in Fig. 5 [in contrast to Fig. 3(b)]. In these structures, boron outdiffusion should also be anticipated, however, since at the $p-n$ junctions, the boron doping exists in layers without carbon. However, as shown in Fig. 5,

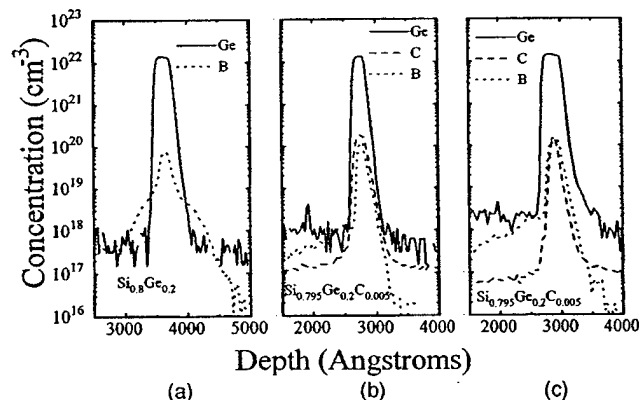


FIG. 4. SIMS profiles of Figs. 1(a), 1(b), and 1(c) (0.5% C) wafers following ion implantation and anneal at 755 °C.

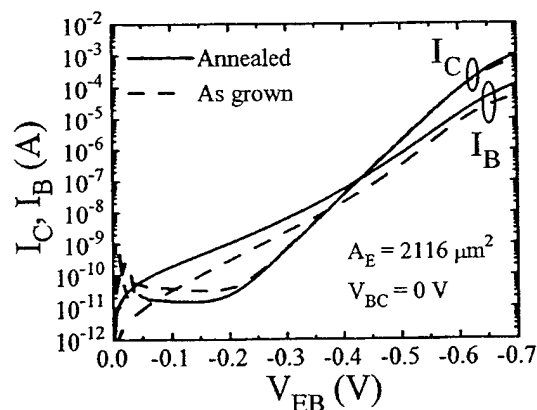


FIG. 5. Gummel plot of wafer of Fig. 4(c) (sandwich base structure) processed both with and without ion implantation and annealing at 742 °C showing no evidence of B diffusion or barrier formation.

the as-grown and annealed (742 °C after As implantation) collector currents are identical, indicating that no outdiffusion has occurred [in contrast to Fig. 3(a)]. Figure 4(c) shows SIMS of the sandwich base following As implantation and anneal, also showing no diffusion. Note that although boron is *outside* the carbon layer, there are no tails due to outdiffusion, as occur in the case for SiGe bases without carbon anywhere. These data suggest that carbon has a nonlocal ability to reduce boron diffusion.

We have shown that carbon in SiGe bases reduces TED-mediated outdiffusion of boron caused by an As-emitter implant and anneal. Other results in our laboratory show a similar reduction in boron diffusion due to thermal annealing without implants and due to TED caused by extrinsic base implants.¹⁰ Previous workers have shown that oxygen levels of 10^{20} cm^{-3} also suppress B diffusion.⁹ Since the oxygen concentration in these bases is low ($\sim 2 \times 10^{18} \text{ cm}^{-3}$) and identical in both wafers with and without carbon, we conclude that carbon is the mechanism that reduces boron diffusion. Carbon in Si has previously been shown to reduce TED of B in silicon due to a Si ion implant and anneal.¹¹ These previous results and our new results suggest that carbon in SiGe is a sink for point defects (presumably interstitials), the mechanism that causes boron diffusion. This would also explain the suppression of boron diffusion in regions nearby the SiGeC, as demonstrated by the sandwiched base struc-

ture. The suppression of boron diffusion by carbon is also accompanied by a decrease in minority carrier lifetime. TEM of all as-grown and implanted and annealed samples shows no evidence of any defects or precipitates in the emitter-base depletion region down to a resolution of $\sim < 20 \text{ Å}$. Therefore, the C-related complex responsible for the sinking of point defects must be very small. It is known that Si interstitials and substitutional carbon can interact in silicon to form small stable complexes on the scale of a few atoms.¹² Further investigation is required to confirm the exact mechanism, however.

In summary, we have shown that the addition of small amounts of C to SiGe layers greatly reduces boron motion due to transient enhanced diffusion. This greatly improves the allowable thermal budgets in Si/SiGe/Si HBT processing and should improve options for the integration of these devices into existing processes.

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Complete suppression of boron transient-enhanced diffusion and oxidation-enhanced diffusion in silicon using localized substitutional carbon incorporation

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In this letter, we show the ability, through introduction of a thin $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ layer, to eliminate the enhancement of enhanced boron diffusion in silicon due to an oxidizing surface or ion implant damage. This reduction of diffusion is accomplished through a low-temperature-grown thin epitaxial $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ layer which completely filters out excess interstitials introduced by oxidation or ion implant damage. We also quantify the oxidation-enhanced diffusion (OED) and transient-enhanced diffusion (TED) dependence on substitutional carbon level, and further report both the observation of carbon TED and OED, and its dependence on carbon levels. © 1998 American Institute of Physics. [S0003-6951(98)00651-2]

As silicon devices are continuously scaled to smaller sizes, there is an ever greater demand to reduce the diffusion of dopants. However, the enhancement of boron diffusion due to ion implant damage and oxidation, transient-enhanced diffusion (TED) and oxidation-enhanced diffusion (OED), poses severe challenges to this goal. It is generally accepted that boron diffuses primarily via an interstitial mechanism and boron diffusivity is, therefore, dependent on the silicon interstitial concentration. Ion implant damage and oxidation are sources of excess interstitials which enhance the boron diffusivity. Therefore, the reduction of TED and OED depend on the effective ability to trap or annihilate any excess interstitials that are produced during processing. Interest in the relationship between carbon and enhanced boron diffusion was recently spurred after TED was observed to strongly depend on the concentration of substitutional carbon in silicon.¹ Various approaches of carbon incorporation into silicon have been used to explore the carbon effect on boron-enhanced diffusion including: carbon containing alloys [e.g., Si_xC_y (Ref. 2) and $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ (Ref. 3)], uniform carbon doping of silicon,¹ and localized carbon doping of silicon.⁴ These approaches may be summarized into two categories. First, silicon or carbon-silicon alloys that have a uniform distribution of carbon in them;^{2,4} and second, silicon with localized carbon-doped regions or thin carbon-containing alloy layers.^{3,4}

Complete suppression of boron TED and OED has already been reported for the category of uniform carbon concentration.⁴ However, for practical use in device applications this approach is hampered by concerns about the electrical activity of carbon defects in silicon,^{3,5,6} and carbon's high diffusivity.

Localized carbon-doped regions, however, offer a potential solution to the difficulties presented by uniform carbon doping. Carbon-containing regions may be placed away from

the active region eliminating concerns about electrical defects, and also creating a test structure with which to isolate carbon's effect on interstitial concentration without the complication of carbon effects on the intrinsic boron diffusivity. However, to the best of our knowledge no complete suppression of TED or OED has been reported to date with a localized carbon region. In this letter we show the ability through introduction of a thin $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ layer to completely filter interstitials injected by an overlying oxidizing surface or interstitials injected due to ion implant damage, which results in the complete elimination of either OED or TED for underlying boron. We also quantify the OED and TED dependence on substitutional carbon level, and further report both the observation of carbon TED and OED, and its dependence on carbon level.

The test structures were grown using rapid thermal chemical vapor deposition (RTCVD),⁷ between 600 and 750 °C using methylsilane as the carbon source. Two boron-doped silicon layers with and without a $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ or $\text{Si}_{1-x}\text{Ge}_x$ layer placed between the peaks were used to test the effect of the layer on boron diffusivities at different locations (above and below) with respect to the SiGe(C) layer. Both boron peaks were approximately 250 Å thick and had a boron concentration of $5 \times 10^{19}/\text{cm}^3$ centered 2000 and 3000 Å away from the surface, respectively, while the 250 Å thick SiGe(C) layer was centered 2300 Å from the surface (Fig. 1).

To study boron OED dependence on substitutional carbon level as-grown samples were cleaved and annealed in nitrogen and oxygen ambient atmospheres at 850 °C for 30 min. Boron profiles were characterized using secondary ion mass spectroscopy (SIMS) with 2 keV Cs ions and an estimated 1%–5% error in depth scales. Figure 2(a) shows profiles of boron in a sample with the SiGe(C) layer omitted, with a background carbon concentration below SIMS detection limits. Broadening of both boron peaks is observed after the nitrogen anneal, and in the oxidation case clear enhancement of the boron diffusion is observed by the further broad-

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i-Si, 1900Å, 700°C
Si : boron $3 \times 10^{19}/\text{cm}^3$, 250Å, 700°C
i-Si, 100Å, 700°C
i-Si _{1-x-y} Ge _x C _y , 250Å, 625°C
i-Si, 400Å, 700°C
Si : boron $3 \times 10^{19}/\text{cm}^3$, 250Å, 700°C
i-Si, 1500Å, 700°C
Si buffer layer : boron $3 \times 10^{19}/\text{cm}^3$, 2500Å, 1000°C

FIG. 1. Test structure to study boron oxidation-enhanced diffusion and transient enhanced diffusion. Two boron markers are separated by a Si_{1-x-y}Ge_xC_y interstitial barrier layer ($x = [0-0.2]$; $y = [0-0.005]$).

ening of the boron profile. The profiles were modeled using a numerical simulator PROPHET, obtained from Lucent technologies, to quantitatively compare boron profiles and obtain boron diffusivities. In this case, the boron diffusivity during oxidation is found to be six times that of the average boron diffusivity during nitrogen anneal. Figure 2(b) shows boron profiles before and after anneals for the sample containing an intervening thin 250 Å Si_{0.795}Ge_{0.2}C_{0.005} (0.5% carbon) layer between the boron markers as indicated by the carbon concentration profile. The same broadening is observed in the peak below the Si_{1-x-y}Ge_xC_y layer after annealing in nitrogen, but in this case there is no additional broadening after oxidation, indicating that for this substitutional carbon level the oxidation-enhanced diffusion has been completely suppressed. As interstitials are injected into the sample at the surface, during oxidation, and there is no observed enhancement of boron diffusion below the Si_{1-x-y}Ge_xC_y alloy layer, we may conclude that the excess interstitial concentration has been reduced to zero below the carbon-containing layer. This demonstrates that the Si_{1-x-y}Ge_xC_y layer acts as an effective interstitial sink, blocking interstitials from reaching the layer below the Si_{1-x-y}Ge_xC_y, and therefore, completely suppressing boron OED.

In addition, for the sample with 0.5% carbon in the in-

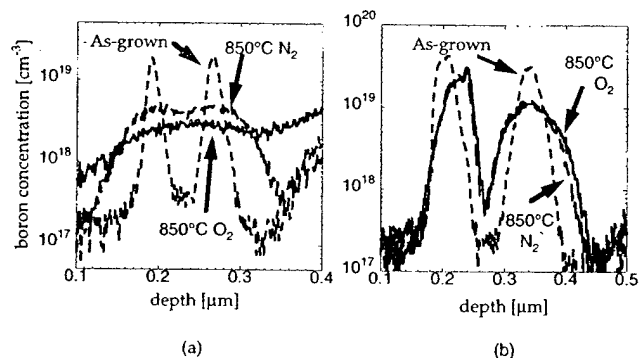


FIG. 2. Boron concentration profiles (measured using SIMS) of as-grown samples, and samples annealed for 850 °C for 30 min in nitrogen ambient, or in oxygen ambient samples. Boron profiles of (a) sample without a SiGeC barrier layer between the boron markers. Oxidation-enhanced diffusion (OED) is demonstrated by the broader boron profile after anneal in oxygen ambient compared to that annealed in nitrogen. Boron profiles of (b) sample with highest level carbon in the SiGeC barrier layer. Boron OED is completely suppressed as indicated by no difference between boron profiles of the deepest marker after anneal in oxygen and anneal in nitrogen ambient.

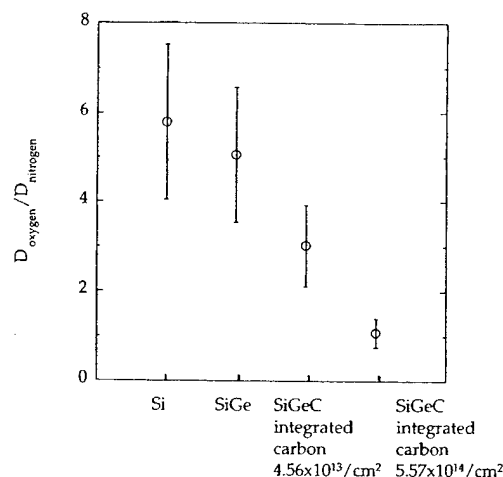


FIG. 3. Relative average boron diffusivity (during oxidation 850 °C for 30 min in O₂) dependence on carbon levels in SiGe(C) barrier layer. Boron diffusivity of deepest marker is displayed relative to that of the average boron diffusivity during anneal in nitrogen ambient, 850 °C for 30 min in nitrogen. Note: oxidation-enhanced diffusion is completely suppressed at highest carbon level.

tervening Si_{1-x-y}Ge_xC_y layer the boron profiles of the shallowest boron marker are nearly identical after anneals in both oxygen and nitrogen [Fig 2(b)], indicating that the excess interstitial concentration must also be reduced to nearly zero even as far as 100–200 Å above the intervening Si_{1-x-y}Ge_xC_y layer. The asymmetric diffusion of the boron in the upper peak may be attributed to boron segregation into the Si_{1-x-y}Ge_xC_y alloy layer. Boron segregation into strained Si_{1-x}Ge_x has been previously reported,^{8,9} and further study is in progress to determine the boron segregation dependence on carbon content.

Samples with different substitutional carbon levels in the Si_{1-x-y}Ge_xC_y alloy layer were grown and annealed to test the OED carbon level dependence. Figure 3 summarizes the relative diffusivities, boron diffusivity of the deepest peak during anneals in oxygen versus nitrogen ambient, for carbon levels of 0%–0.5%. To within experimental error there is no difference between the OED of boron in the cases of boron markers in pure silicon and that with an intervening Si_{1-x}Ge_x alloy layer (no carbon) between the boron markers. This demonstrates that the interstitial trapping is not a result of the addition of the Si_{1-x}Ge_x alloy layer and is in agreement with previous work.^{3,10} Finally, we observe a monotonic decrease in boron OED with increasing carbon level in the alloy layer and the total carbon level required to completely suppress the OED for these conditions is bound between the 4.56×10^{13} and $5.57 \times 10^{14}/\text{cm}^2$ carbon level.

To the dependence of boron TED on the substitutional carbon level, as-grown samples (Fig. 1) with and without buried Si_{1-x-y}Ge_xC_y layers were cleaved and subjected to implant levels of 5×10^{13} and $5 \times 10^{14}/\text{cm}^2$ silicon with implant energies of 30 keV, then annealed in nitrogen ambient for 15 min at 800 °C. The depth of the silicon implanted profile is estimated to be 600 Å. Migration of interstitials from the ion implant damaged region leads to excess diffusion of the lower boron peak (as measured by SIMS), which decreases as the carbon level of the barrier layer is increased. At the highest carbon levels no significant difference in the lower boron profiles can be seen between samples with or

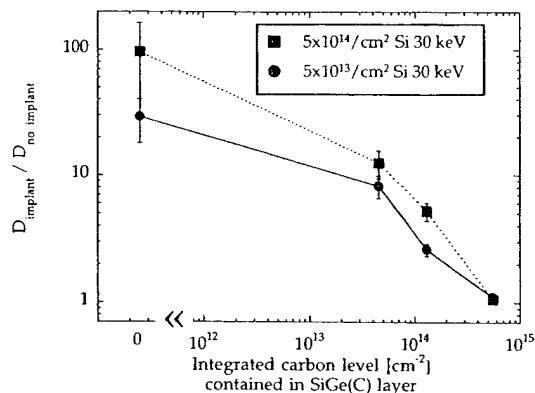


FIG. 4. Average boron diffusivity (during 800 °C 15 min anneal after ion implantation for Si doses of 5×10^{13} or $5 \times 10^{14}/\text{cm}^2$ at 30 keV) dependence on carbon levels in SiGe(C) barrier layer. Boron diffusivities of deepest marker is displayed relative to that of the average boron diffusivity during an 800 °C 15 min anneal in nitrogen ambient atmosphere without ion implantation. Note: transient-enhanced diffusion is completely suppressed at highest carbon level.

without ion implant damage, again indicating complete filtering of the excess interstitials. Average boron diffusion constants were then extracted by fitting measured boron profiles to simulation. Figure 4 summarizes, for different carbon levels in the barrier level, the average boron diffusivity after ion implantation, for two different silicon implant doses, relative to the measured boron diffusivity without silicon ion implantation in the same samples. For both implant doses of 5×10^{13} and $5 \times 10^{14}/\text{cm}^2$ silicon the total carbon level required to completely suppress the TED for these conditions is bound between the 4.56×10^{13} and $5.57 \times 10^{14}/\text{cm}^2$ carbon level.

Because carbon is electrically active it is important to determine where carbon is after processing. Carbon is reported to diffuse quickly in silicon.⁶ Figure 5(a) shows the carbon profiles of as grown, annealed in nitrogen, and annealed in oxygen for a sample with an intermediate carbon level ($4.6 \times 10^{13}/\text{cm}^2$). Comparison of carbon profiles after annealing in oxygen versus nitrogen shows an enhanced diffusion in the oxygen case. This carbon OED is, however, completely suppressed in samples with higher carbon levels, Fig. 5(b), where no difference can be seen in the carbon profiles between anneals in nitrogen and oxygen. Similar carbon diffusivity dependence on total carbon level is also observed in TED samples.

In summary, boron OED and TED in silicon has been completely suppressed by trapping excess interstitials in an overlying $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ layer. Carbon diffusion is enhanced during oxidation, however, at the highest carbon levels this

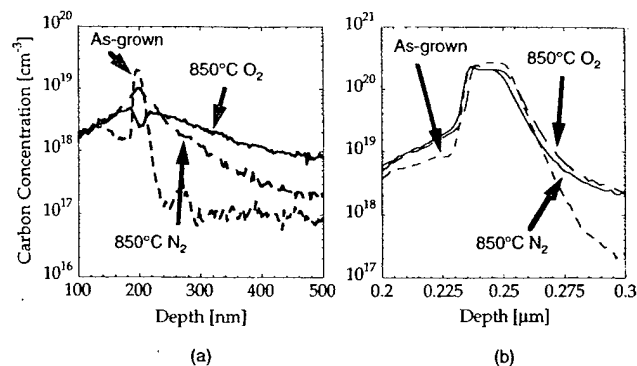


FIG. 5. Carbon concentration profiles (measured using SIMS) of samples that were annealed for 850 °C for 30 min in oxygen ambient. Carbon profiles of (a) samples with the lowest level of carbon used in the SiGe(C) barrier layer between boron markers. Carbon oxidation-enhanced diffusion demonstrated in this sample by the additional broadening of the carbon profile after anneal in oxygen ambient compared to that annealed in nitrogen. Carbon profiles of (b) samples with highest level carbon in the SiGe(C) barrier layer. Carbon OED completely suppressed as indicated by no difference between carbon profiles after anneal in oxygen and anneal in nitrogen ambient.

enhancement was completely suppressed so that the boron-doped silicon region remains free of carbon under conditions when boron OED/TED is also suppressed. Due to recombination of interstitials at the surface and other uncertainties, further investigation is still needed to determine the relationship between carbon level and silicon interstitials, in order to effectively model carbon's effect on boron and carbon diffusion.

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13. ABSTRACT (Maximum 200 words) The main overall goal of this project was to improve the high speed performance of transistors and integrated circuits based on silicon, which is the material on which most microelectronic circuitry (such as microprocessors and memory chips) is based. More specifically, we addressed the three-fold gap (150 GHz vs. 50 GHz top speeds) between the performance of heterojunction bipolar transistors made using silicon-germanium technology in isolated devices in research labs vs. that achieved in integrated circuits in production. We have shown that the incorporation of small amounts of carbon into the silicon-germanium can overcome one of the main problems which lead to the much lower performance of integrated devices vs. that of isolated laboratory devices. The carbon is effective because it greatly reduces the unwanted movement of atoms during the microfabrication process.				
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